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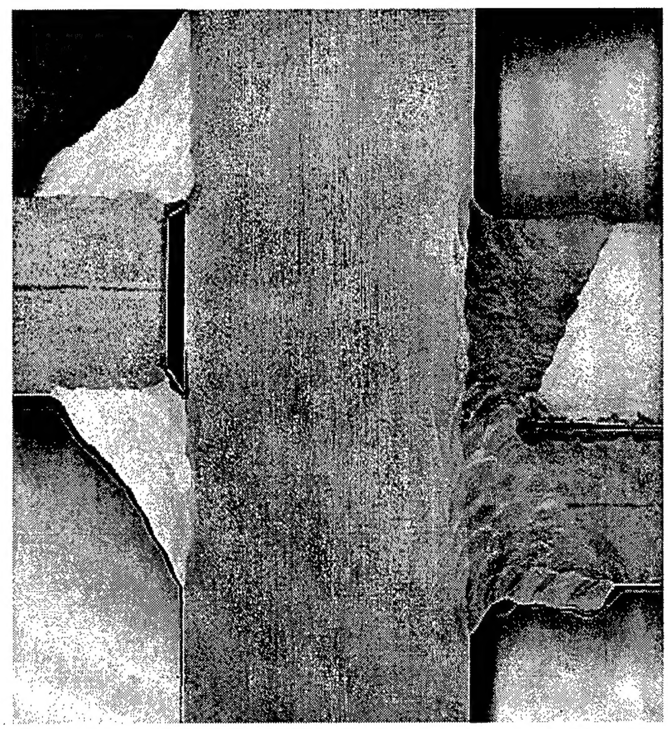
Low-Carbon, Age-Hardenable Steels for Use in Construction: A Review

by
Robert A. Weber, Bruce R. Somers, and Eric J. Kaufmann

High-strength low-alloy (HSLA) steels are being used extensively in U.S. Naval shipbuilding and are being substituted for quenched and tempered HY-80 steel plate. This report examines other applications for the use of HSLA steels. They were studied to determine possible uses in U.S. Army Corps of Engineers construction projects, the current technology gaps, and the research necessary to fill those gaps.

HSLA steels offer some benefits over conventional steels of the same strength level. They are virtually immune to hydrogen-assisted cracking in the heat affected zone of welds, which allows HSLA steel to be welded without preheating. However, at higher strength levels the weld metals used may require preheat to prevent weld metal hydrogen-assisted cracking.

The low-carbon, fine grained microstructure that results from typical processing yields a favorable combination of excellent fabricability, strength, and toughness to HSLA steel that adds to its usefulness and gives it clear advantages over quenched and tempered construction steels. However, fatigue and buckling may limit direct design substitutions of HSLA steel in Corps new construction application. HSLA steel is resistant to hydrogen-assisted cracking but susceptible to reheat cracking, and applications requiring post weld heat treatment are not recommended. Also, local brittle zones may result in low toughness.



Cross-section of full penetration beam-to-column weld with beam flange stiffener plate.

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Foreword

This research was conducted for the Directorate of Military Programs, Headquarters, U.S. Army Corps of Engineers (HQUSACE), under Project 4A162784AT41, "Military Facilities Engineering Technology"; Task MA, Work Unit CB2, "Assessment of HSLA Steels for Construction." The technical monitor was Charles Gutberlet, CEMP-ET.

The work was performed by the Materials Science and Technology Division (FL-M) of the Facilities Technology Laboratory (FL), U.S. Army Construction Engineering Research Laboratories (USACERL). The USACERL principal investigator was Robert A. Weber. Bruce R. Somers and Eric J. Kaufmann are researchers with the Center for Advanced Technology for Large Steel Structures, Lehigh University, Bethlehem, PA. Ilker R. Adiguzel is Acting Chief, CECER-FL-M, and Donald F. Fournier is Acting Operations Chief, CECER-FL. The technical editor was Agnes E. Dillon, Technical Information Team.

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COL James T. Scott is Commander of USACERL, and Dr. Michael J. O'Connor is Director.

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1 Introduction

Background

Over the past decade, ASTM A710-type, low-carbon, age-hardenable steels have been used increasingly in manufacturing and construction because of their excellent weldability and fracture toughness. The largest tonnage of this type of steel probably is used in U.S. Naval shipbuilding, and these steels are covered by Military Specification MIL-S-24645, which includes both an 80 kips per square inch (ksi) and a 100 ksi minimum yield strength material. Shipbuilding steels have become known as high-strength low-alloy (HSLA) steels, although they are not truly within the confines of the conventional definition of HSLA steels because their total alloy content generally is about 4 percent.

The use of microalloyed steels for construction and pressure vessel applications goes back to the 1940s when critical material shortages made the use of only small amounts of alloys in steel products mandatory. In the 1950s, a number of companies introduced construction steels that contained manganese and other alloys as well as vanadium, titanium, and niobium. Most of these steels are now covered by ASTM A588 and A572, and are considered the early microalloyed steels, although they were not originally identified as such. In the 1970s microalloyed steels with low carbon content, high manganese levels, and microalloy carbide and nitride formers became identified as construction materials with high strength, good weldability, and good low temperature toughness. These materials allow control of grain size and microstructure, either as-rolled or specially processed, to provide a good combination of properties for a number of applications.

Some of the first uses of intentional additions of copper to steels were provoked by the appreciable atmospheric-corrosion resistance developed. The ASTM A242 and A588 steels possess at least four to as much as eight times the atmospheric-corrosion resistance of structural carbon steel, which has a low copper content. When properly exposed to the atmosphere, such bare steels will develop a tight, adherent, protective-oxide coating during the first several months of weathering. Thereafter, little additional steel corrosion occurs.

The use of copper in amounts over 0.2 percent to strengthen structural steels also goes back to at least the 1940s, when it was used in some ships. Structural steels with copper additions for precipitation strengthening, forerunners of the current A710-type grades, were introduced in the 1960s but generated little interest among engineers and the construction industry at that time. However, engineering needs change, and now there is considerable interest in the combination of properties that copper age-hardening steels of the A710 type provide.

The current interest in these materials by the U.S. Army Corps of Engineers (USACE) raises questions about the utility of the materials for Corps construction projects. Can this material reduce the section thicknesses required for common construction applications and still provide the required service life? This and similar types of questions prompted an investigation of the A710 steels for possible application in Corps construction.

Objectives

The objectives of this work were to study the high-strength low-alloy ASTM A710-type steels to determine possible uses in USACE construction projects, the current technology gaps, and the research necessary to fill those gaps.

Approach

This report comprises information compiled from the technical literature and interviews with fabricators, steelmakers, and researchers involved with the use, supply, and evaluation of HSLA steels.

Scope

The scope of this work was limited to HSLA steels and their application to the construction of civil structures typical of the U.S. Army Corps of Engineers military construction programs.

Mode of Technology Transfer

The pertinent information in this report will be incorporated in the next update of the Technical Manual TM5-805-7, *Welding: Design, Procedures, and Inspection*.

Metric Conversion Factors

U.S. standard units of measure are used throughout this report. A table of metric conversion factors is presented below.

1 in.	=	25.4 mm
1 ft	=	0.305 m
1 sq ft	=	0.093 m^2
1 sq ft/min	=	0.093 m^2/min
1 cu ft	=	0.028 m^3
1 mi	=	1.61 km
1 lb	=	0.453 kg
1 gal	=	3.78 L
1 psi	=	6.89 kPa
1 µm	=	1×10^{-6}m
°F	=	(°C × 1.8) +32

2 High-Strength Low-Alloy Materials

Metallurgy

Low-carbon, copper precipitation-aged plate steels were introduced in the late 1960s by the International Nickel Company as IN-787. The current ASTM specification (A710) for structural applications was developed more than 10 years ago. For pressure vessel applications, this material is covered by ASTM specification A736. Both A710 and A736 steels can be supplied in three different classes: Class 1, as-rolled; Class 2, normalized; and Class 3, quenched. All three classes are precipitation heat treated at 1000 to 1300 °F (540 to 705 °C) for 30 to 60 minutes. They exhibit a wide range of tensile strength, 65 to 120 ksi (450 to 650 MPa) as well as good impact toughness at low temperatures.

Table 1 lists the compositions of the three ASTM grades of these A710-type steels and the HSLA-80 and HSLA-100 of the MIL-S-24645. These are extremely low carbon steels (<0.07 percent carbon) with moderate alloy additions of manganese, nickel, molybdenum, and chromium and with significant copper, which provides the age-hardening characteristic.

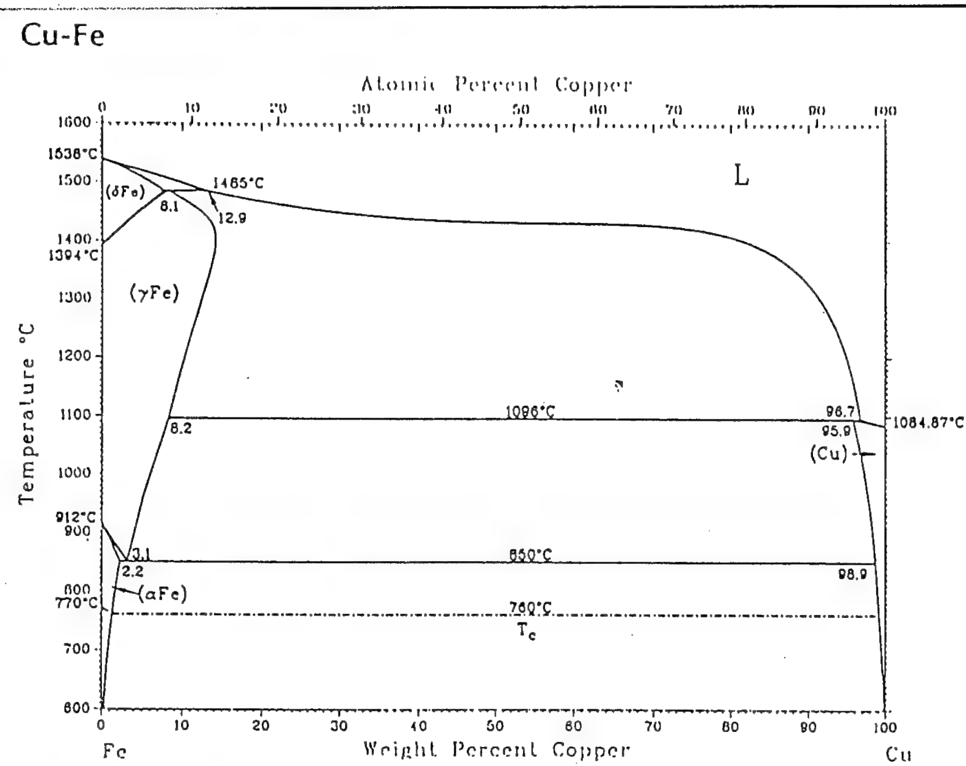
Copper has a high solubility in austenite at typical austenitizing temperatures, 1600 to 1750 °F (870 to 950 °C), as shown in Figure 1. Ferrite formed at high temperature also has a relatively high solubility for copper, slightly over 2 percent maximum. However, due to the sloping solvus line, the copper solubility drops significantly at lower temperatures. Thus, a rapidly cooled alloy contains copper in a supersaturated condition. When reheated, copper-rich precipitates form as fine spherical particles and cause the precipitation-hardening of the steel. These particles usually are referred to as epsilon-copper. Steels containing more than 0.60 percent of copper are capable of exhibiting precipitation hardening of the ferrite.

Copper is known to have the potential to cause hot shortness problems due to molten iron-copper phases at the scale interface during heating in oxidizing atmospheres (Jesseman and Murphy, June 1984). The addition of nickel prevents this problem.

Table 1. Chemical composition requirements of low-carbon age-hardenable alloy.

	ASTM* A710/736	ASTM A710	ASTM A710/736	MIL-S-24645**	
	Grade A	Grade B	Grade C	HSLA-80	HSLA-100
Carbon	0.07	0.06	0.07	0.06	0.06
Manganese	0.40-0.70	0.40-0.65	1.3-1.65	0.40-0.70	0.75-1.05
Phosphorus	0.025	0.025	0.025	0.020	0.020
Sulfur	0.025	0.025	0.025	0.006	0.006
Silicon	0.40	0.15-0.40	0.40	0.40	0.40
Nickel	0.07	0.06	0.07	0.06	0.06
Chromium	0.60-0.90	---	---	0.60-0.90	0.45-0.65
Molybdenum	0.15-0.25	---	0.15-0.25	0.15-0.25	0.55-0.65
Copper	1.00-1.30	1.00-1.30	1.00-1.30	1.00-1.30	1.45-1.75
Niobium	0.02 min	0.02 min	0.02 min	0.02-0.06	0.02-0.06

* ASTM specifications.
**Military specifications.



(Source: Baker 1992).

Figure 1. Copper-iron equilibrium phase diagram.

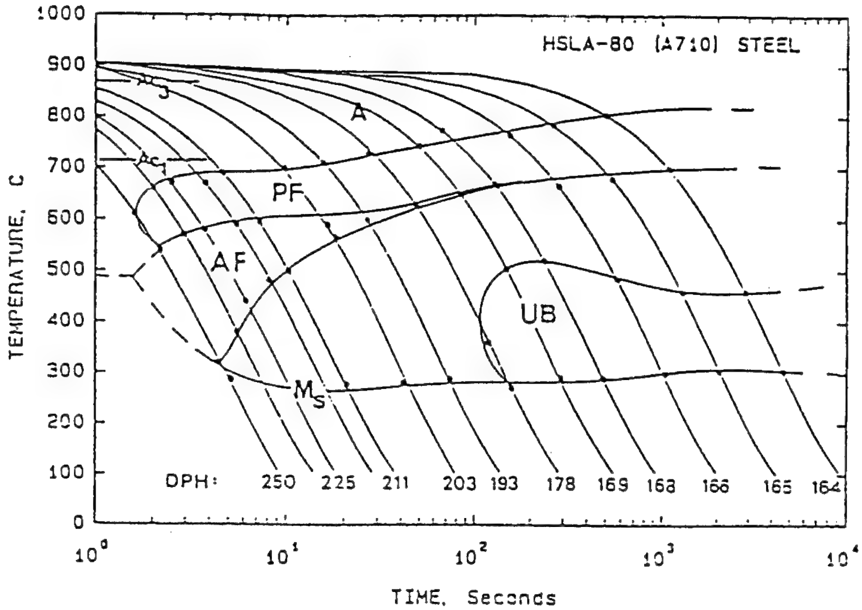
Nickel also provides toughness. Chromium and molybdenum are added to control the epsilon-copper precipitate nucleation and growth so consistent properties can be developed. Chromium and molybdenum also provide additional hardenability that helps promote a fine ferritic/bainitic/martensitic microstructure. Niobium helps retard austenite recrystallization during hot rolling and makes grain refinement possible. Niobium also provides some precipitation hardening. Grain size control during austenitizing is provided by the niobium carbo-nitride precipitates.

Figure 2 shows continuous cooling transformation (CCT) diagrams for an A710 composition and an HSLA-100 composition (Wilson, et al., September 1988). The increased hardenability of the HSLA-100 alloy is evidenced by the displacement of the proeutectoid ferrite field to longer times or slower cooling. The A710 CCT diagram consists of proeutectoid ferrite, acicular ferrite, martensite, and upper bainite. Figure 2B shows an extensive martensitic region, with granular bainite and proeutectoid ferrite at slower cooling rates. Granular bainite contains packets of ferrite laths with noncementite, interlath, second-phase particles. These particles have been identified as retained austenite or a combination of retained austenite and martensite (Hamburg and Wilson, November 1987). These structures are difficult to distinguish from upper bainite, which consist of ferrite and cementite. Granular bainite occasionally is observed in A710 compositions (Wilson et al., September 1988).

Table 2 summarizes the mechanical properties available via various processing conditions as allowed by ASTM and military specifications. The ASTM materials can be furnished in one of three conditions or classes as they are termed in the specifications.

Class I—As-rolled and Precipitation Heat Treated. The ferritic microstructures formed by conventional hot-rolling or controlled rolling typically have yield strengths from 65 to 75 ksi (450 to 525 MPa). Reheating in the range of 1000 to 1200 °F (540 to 650 °C) causes the supersaturated copper to precipitate and raises strengths. Controlled rolling is often used to achieve adequate toughness in thicker Class 1 plates. The thickness limits are based on mill capacity. The rolling mill must introduce sufficient strain at low rolling temperatures to achieve grain refinement. Currently, the maximum Class 1 plate thickness in ASTM is 3/4 in. (19 mm) and 1/2 in. (13 mm) in the military specification.

Class 2—Normalized and Precipitation Heat Treated. This process requires a normalization at 1600 to 1700 °F (870 to 930 °C) followed by reheating at 1000 to 1200 °F (540 to 650 °C) to precipitate epsilon-copper in the ferrite and raise strengths.



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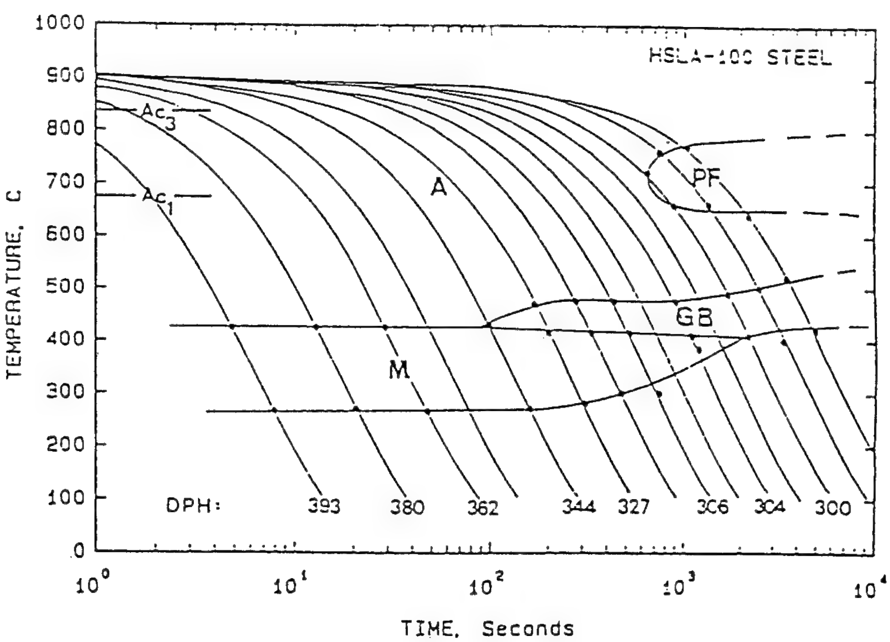
A = Austenite, PF = Proeutectic Ferrite, AF = Acicular Ferrite, UB = Upper Bainite, M_s = Martensite Start.



b

Figure 2. Continuous cooling transformation diagrams.

Table 2. Mechanical property requirements per ASTM: low-carbon age-hardenable alloys.

Specification	Grade condition*	Yield strength (ksi)	Tensile strength (ksi)	%E	Charpy V-notch
A710	Grade A Class 1	85, t<5/16" 80, 5/16<t<3/4"	90, t<3/4"	20	20ft-lb@-50F, Long.** 15ft-lb@-50F, tran
	Grade A Class 2	65, t<1" 60, 1"<t<2" 55, 2"<t<4" 50, t>4"	72, t<2" 65, 2"<t<4" 60, t>4"	20	50ft-lb@-50F***
	Grade A Class 3	75,t<2" 65, 2"<t<4" 60, t>4"	85, t<2" 75, 2"<t<4" 70, t>4"	20	50ft-lb@-80F***
	Grade B	85, t<5/16" 82, 5/16<t<3/8" 80, 3/8<t<1/2" 75, 1/2<t3/4"	90, t<1/2" 88, 1/2<t<3/4"	18	20ft-lb@-25F, long 15ft-lb@-25F, tran***
	Grade C Class 1	90,t<3/4"	100, t<3/4"	20	20ft-lb@-50F, long 15ft-lb@-50F, tran**
	Grade C Class 3	85, t<3/4" 80, 3/4"<t<2"	95, t<1/2" 90, 1/2"<t<2"	20	50ft-lb@-80F***
A736	Grade A Class 1	80, t<3/4"	90-110, t<3/4"	20	20ft-lb@-50F, long
	Grade A Class 2	65, t<1" 60, 1"<2" 55, 2"<t<4" 50, t>4"	72-92, t<2" 65-85, 2"<t<4" 60-80, t>4"	20	20ft-lb@-50F, long
	Grade A Class 3	75, t<2" 65, 2"<t<4" 60, t>4"	85-105, t<2" 75-95, 2"<t<4" 70-90, t>4"	20	20ft-lb@-50F, long
	Grade C Class 1	90, t<3/4"	100-120, t<3/4"	20	20ft-lb@-50F, long
	Grade C Class 3	85, t<3/4" 80, 3/4<t<2"	95-115, t<3/4" 90-110, 3/4"<t<2"	20	20ft-lb@-50F, long
MIL-S 24645	HSLA-80	80-100, t<1.25"	110, t<1/2" Info only, t>1/2"	14,t<1/4" 20,t>1/4"	60ft-lb@-120F 35% shear@-120F
	HSLA-100	100-130, t<3/4" 100-125, 3/4<t<4"	Info only	12,t<1/4" 17,1/4<t<3/4" 18,3/4<t<4"	80ft-lb@-80F 90% shear@-80F 60ft-lb@-120F 35% shear@-120F

*Class 1, as-rolled and precipitation heat treated; Class 2, normalized and precipitation heat treated; Class 3, quenched and precipitation heat treated.
**On agreement, transverse tests may be specified instead of longitudinal.
***Optional supplementary test requirements.

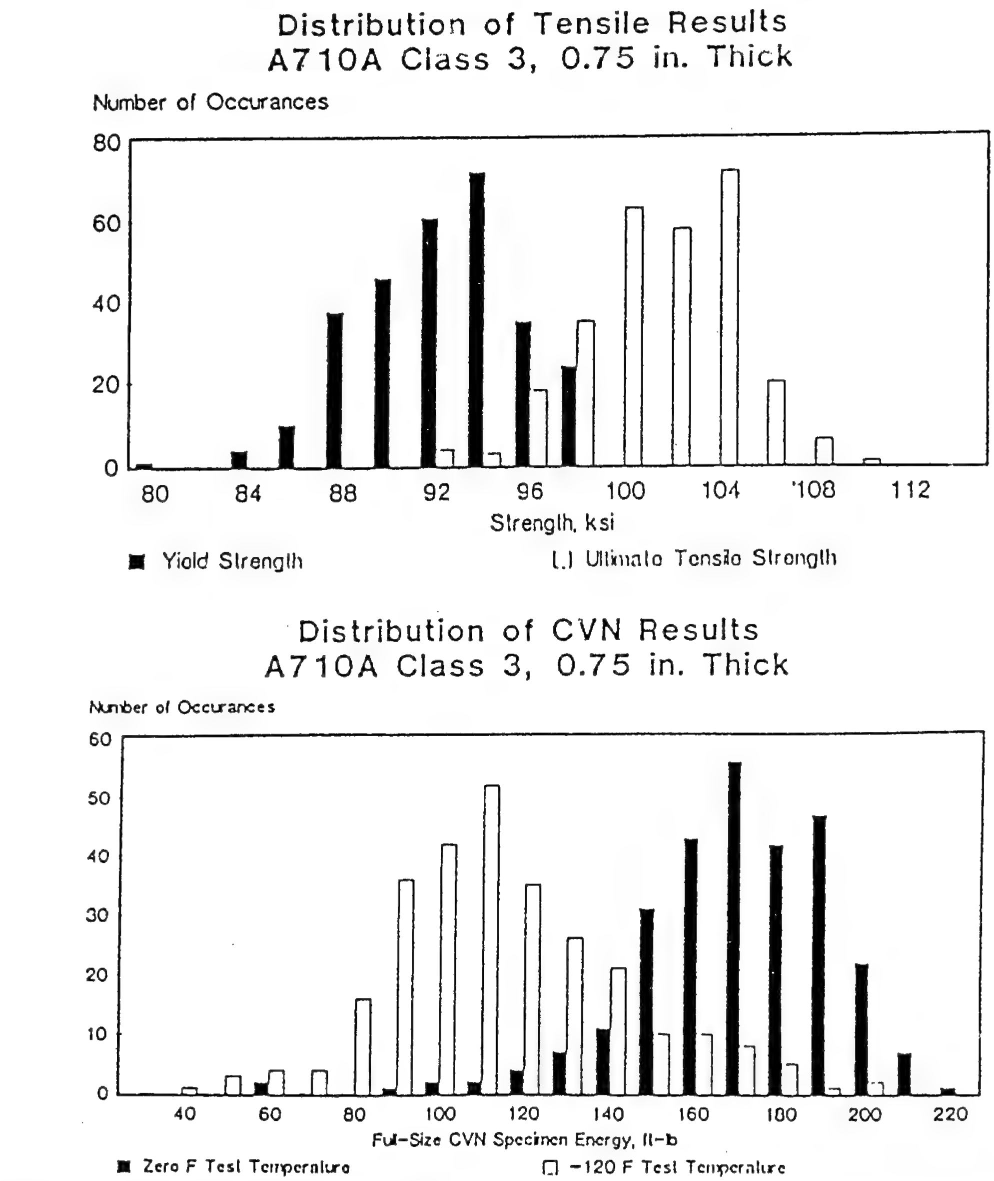
Class 3—Quenched and Precipitation Heat Treated. Austenitizing at 1600 to 1700 °F (870 to 930 °C) and water quenching provides a fine supersaturated microstructure with a yield strength of about 65 ksi (450 MPa). Precipitation hardening at 1000 to 1200 °F (540 to 650 °C) causes copper-rich precipitates to form. The strength increase provided by this age hardening more than offsets the microstructural softenng (tempering) that occurs simultaneously at the precipitation heat treatment temperatures. To a certain extent, a tempering reaction, precipitation, and/or growth of carbides may be seen in all three classes; however, it is more pronounced in Class 3 because the more rapid cooling develops a finer microstructure in a lower temperature transformation product.

Except for the lower allowable sulfur level in the military specification, the military specification HSLA-80 composition is essentially identical to Grade A of the ASTM specification. This low sulfur level is achieved by calcium treatment, and generally the steel makers supply both the military and the ASTM materials with this treatment. The HSLA-100 grade is supplied only in the Class 3 condition. The military specification allows for HSLA-80 up to 1/2 in. thick to be supplied in the Class 1 condition.

Figure 3 shows the range of yield and tensile strengths and Charpy V-notch (CVN) results typical from production heats of A710. These materials are capable of superior combinations of strength and toughness, with yield strengths nearly 100 ksi and -120 °F CVN values averaging over 100 foot-pounds (ft·lb). When more stringent dynamic tear toughness requirements are imposed, controlled rolling prior to reaustenitizing, quenching, and aging has been found to be helpful in achieving additional toughness (Hamburg and Wilson, November 1987).

The CCT diagram for A710 steels (Figure 2B), shows that a range of microstructural constituents can be formed in this low carbon steel; but in practice, proeutectoid ferrite is the primary microstructural component. Small amounts of acicular ferrite, bainite, and martensite also might be formed. The carbon content of A710 steel is typically restricted to less than 0.07 percent so high toughness levels can be developed, and cold cracking problems essentially can be eliminated, even if the welding consumables can generate a hydrogen potential.

Typically, because of the limited hardenability of the A710 composition, the plate thickness is restricted to 1.25 in. for high toughness applications. However, thicknesses up to 5 in. are possible if only modest toughness is required (Wilson et al., September 1988). A modified A710 with higher manganese (1.4 percent) and molybdenum (0.45 percent) content has been found capable of achieving 80 ksi yield

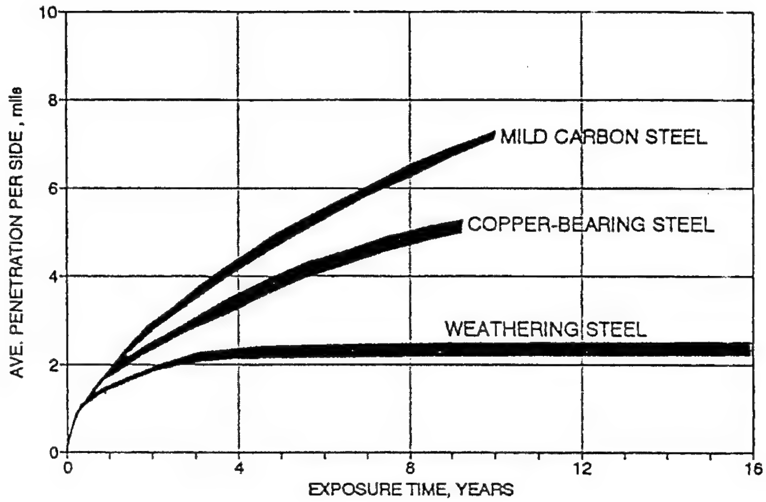


(Source: Wilson et al., September 1988).

Figure 3. Distribution of properties in 0.75 in. (19 mm) thick HSLA-80/A710 Grade A Class 3 plates.

strength up to 8 in. thick with a modest toughness level, 73 ft·lbs at -50 °F (Wilson et al., September 1988).

Because of the copper content, can these A710-type steels exhibit good atmospheric corrosion resistance–weathering behavior–similar to that of the ASTM A588 and A242 weathering steels? These weathering steels develop a tightly adherent, protective rust after 3 to 4 years of wet-dry atmospheric exposure. This behavior stems from a carefully balanced composition with typically 0.35 percent copper, 0.50 percent chromium, and 0.50 percent nickel. Figure 4 shows a comparison of the



(Adapted from Horton, not dated).

Figure 4. Comparison of atmospheric corrosion behavior of several steel types.

corrosion penetration of several steels exposed to an industrial environment (Horton, not dated). Corrosion nearly stops after 3 years in the weathering steel; however, corrosion continues in both the plain carbon and copper-bearing steel, although the copper-bearing steels corrode at about half the rate of the plain carbon steels.

Unfortunately little data is available on the atmospheric corrosion behavior of the A710 type steels. Although their compositions are similar to weathering steels, it is not clear how the higher copper and finer microstructure affect the A710 steels' corrosion behavior. However, most steelmakers agree that the corrosion rates should be a least similar to the copper-bearing steels. Corrosion research has been performed for the U.S. Navy and has concentrated on marine environments and stress corrosion cracking susceptibility. These results indicate these steels have good resistance to stress-corrosion cracking in marine environments (Aylor et al., May 1990).

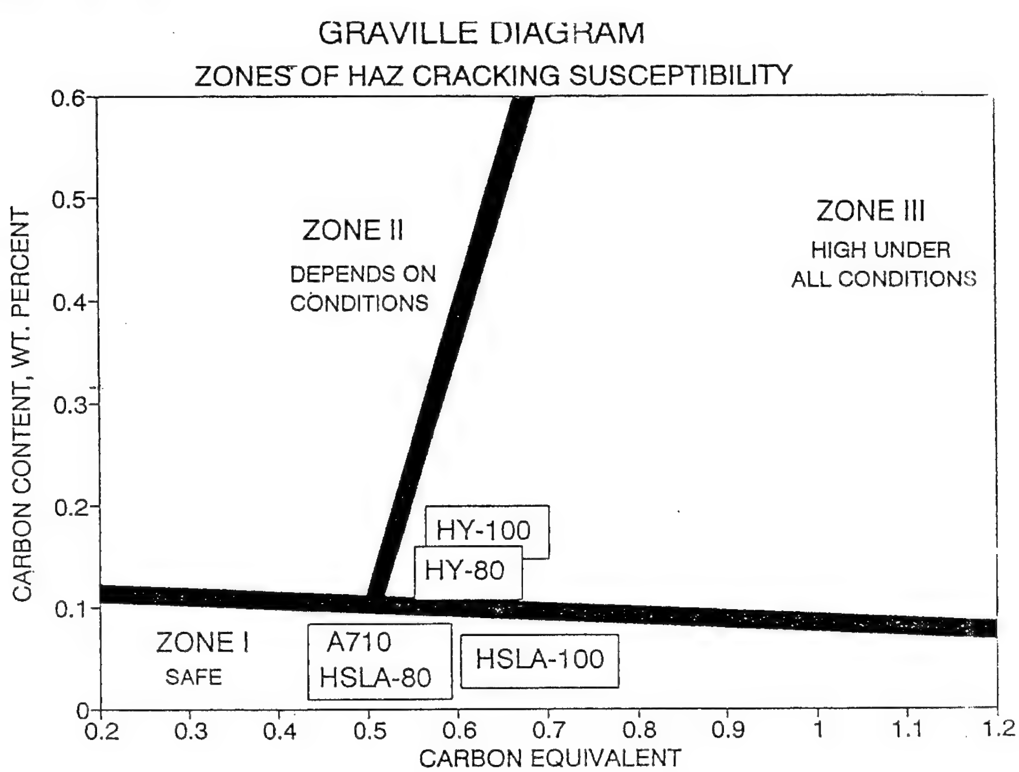
Weldability

The weldability of the HSLA steels has been extensively studied over the past 10 years (Wilson, March 1987; Lundin, Menon, and Lawson, December 1989; Kvidahl, July 1985; Deb, Challenger, and Burna, November 1985; Wallace and Heid,

November 1990; Bollinger et al., September 1988; Balaguer, Wang, and Nippes, April 1989; Kim and Choo, September 1988; Lundin et al., December 1990; Abe, Kurihara, and Tagawa, September 1988; Scoonover, November 1990; Kapadia et al. unpublished; Churchill, Devletian, and Singh, May 1991; Stuart, January 1991; West, May 1987; Anderson, Hyatt, and West, September 1987; Kvidahl unpublished; Castner et al., April 1993; Dexter, Fisher, and Beach, unpublished). A710-type steels are readily weldable under almost all shop and field conditions using a wide variety of welding processes and procedures. Generally no preheat is required for these materials, primarily because of the very low carbon content.

Hydrogen-Assisted Cracking

The potential for excellent weldability in HSLA steels is illustrated in Figure 5. This diagram, proposed by Graville (Wilson et al., September 1988), relates susceptibility to hydrogen-assisted cold cracking to carbon content and carbon equivalent. The A710-type of materials are in zone I of the Graville diagram because of their low carbon content, thus they are expected to be relatively hydrogen-crack free under most welding conditions. Studies have shown that A710



$$CE = C+(Mn+Si)/6+(Ni+Cu)/15+(Cr+Mo+V)/5$$

(Source: Wilson et al., September 1988).

Figure 5. Weldability of steel as a function of carbon content and carbon equivalent.

steel has a very low potential for martensite formation in the grain-coarsened heat affected zone (Lundin, Menon, and Lawson, December 1989). A710 steel is expected to posses superior resistance to hydrogen-assisted cracking. As a testament to this excellent weldability, one U.S. Navy investigation conducted more than 100 weldability tests over a range of processes and heat inputs, no test showed any heat-affected zone (HAZ) cracking (Kvidahl, July 1985).

There have been some concerns for the cracking resistance of the weld metal (Deb, Challenger, and Burna, November 1985; Wallace and Heid, November 1990), especially when matching strength in the higher strength modifications of these A710-type materials. If preheat is not used, the hydrogen cracking problem may move to the weld metal portion of the joint when the strength levels begin to approach 100 ksi, this is particularly true for the processes that are more prone to hydrogen pick-up. Tests at Newport News Shipyard have suggested that a preheat of 125 °F (51.6 °C) minimum is necessary to avoid weld metal cracking in HSLA-100 weldments over 3/4-in. (19-mm) thick welded with matching strength shielded metal arc electrodes (Wallace and Heid, November 1990).

Reheat Cracking

Another area of concern in the HSLA alloys is their susceptibility to reheat cracking. Reheat cracking, also referred to as stress-relief cracking (SRC) and post weld heat treatment (PWHT) cracking, is defined as cracking that occurs in the HAZ during the exposure of welded assemblies to post weld heat treatment (PWHT) or high temperature service. This type of cracking has been found to be predominantly located in the coarse-grained HAZ and is intergranular in nature (Balaguer, Wang, and Nippes, April 1989).

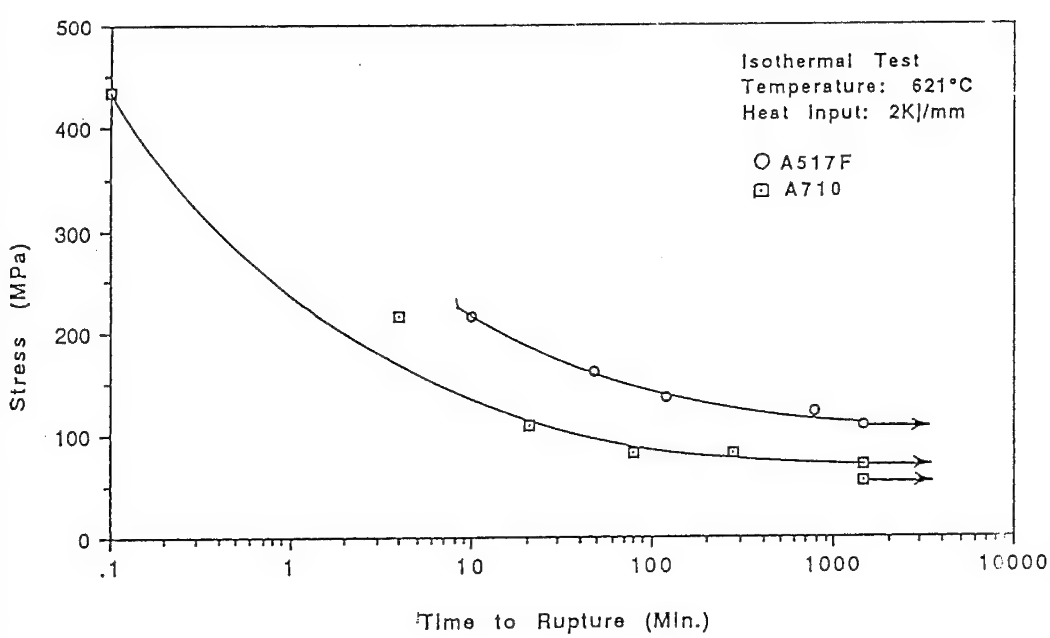
Meitzner (November 1975), in a comprehensive review of SRC in weldments, stated that SRC was associated with: (1) precipitation-hardening materials, (2) intergranular failure with little or no evidence of deformation, and (3) the coarse grained region of weld HAZ. Most SRC cracking has been associated with alloy systems that undergo precipitation hardening, e.g., some low-alloy steels, ferritic creep-resistant steels, austenitic stainless steels, and nickel-based alloys.

Although the exact mechanism has not been clearly established, the basic characteristics of cracking are similar for all the HSLA materials. No external stress is necessary, the crack driving force is derived from residual stress in sound weldments that are heated to elevated temperatures after welding. Therefore, cracks most often are found in thicker sections with high levels of self-restraint, but they may occur even in sheet details if the restraint from fabrication is high enough. The HAZ

cracks are intergranular in nature, typical of high-temperature, stress-rupture failures. The failed specimens exhibit low ductility, and show little or no evidence of deformation. Reheat cracks usually are confined to the coarse-grain region of the HAZ, thus cracking is directly related to prior exposure to extremely high temperatures.

Evaluations of the susceptibility of the A710-type materials to reheat cracking have shown the A710 steels to be one of the most vulnerable of all steels known to be prone to this type of cracking (D. Chen, E.J. Kaufman, B.R. Somers, and A.W. Pense, unpublished data) Figure 6 shows the reheat cracking response of an A710 plate compared to an A517F (T-1) plate, a material known to be quite susceptible to reheat cracking. Figure 6 shows a low critical stress for reheat cracking for this A517F plate, 107 MPa (15.3 ksi), but an even lower critical stress for the A710 plate, 74 MPa (10.7 ksi).

One major service failure due to this type of cracking is known to the authors. It involved a large vessel used to contain molten metal. The vessel was fabricated from a thick A710 alloy plate. The cracking occurred because of exposure to an elevated service temperature. Fortunately the cracking was discovered before a major catastrophe occurred.



(Source: Balaguer, Wang, and Nippes, April 1989.)

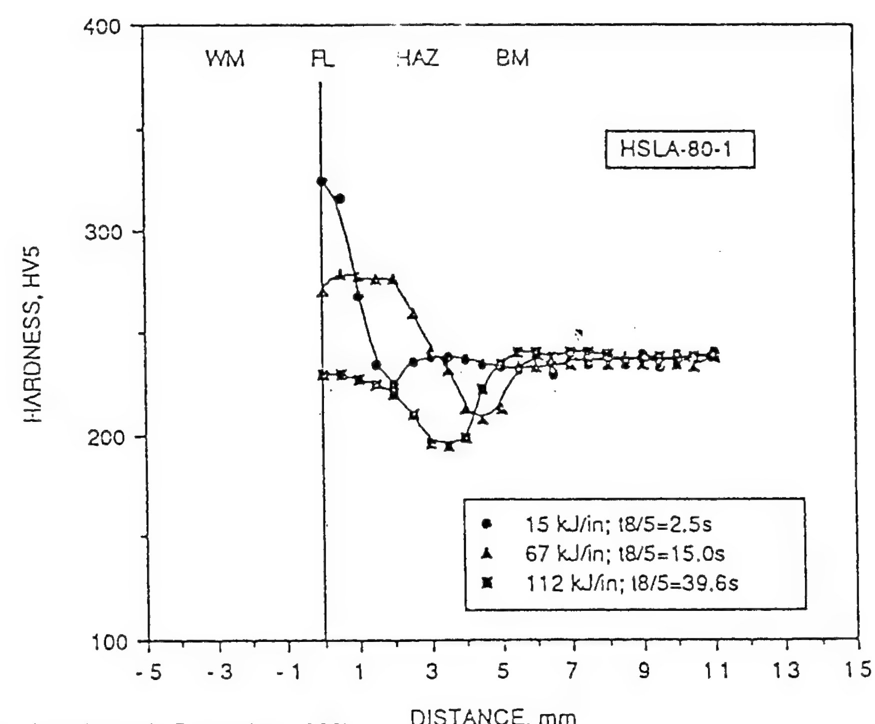
Figure 6. Comparison of reheat cracking susceptibility in A710 and A517F as revealed by the stress rupture implant test at 1150 °F (621 °C).

Recently a railcar designer/fabricator considered fabricating a railcar truck from A710 steel plate. As part of the evaluation, because of the close tolerances required for the axles and other machine parts, it was determined to be necessary to PWHT the assembly. Subsequent studies of the reheat cracking susceptibility of A710 steels caused the company to reject these steels as a candidate material for railcar trucks.

Heat-Affected Zone Softening

Because HSLA steels are age-hardened, there always is the possibility that subsequent thermal cycles, such as experienced during welding, can change the size and distribution of strengthening precipitates. Thus in the HAZ of a weld one might expect to find zones in which: (1) the epsilon-copper has been dissolved and the copper put back into a supersaturated solution, or (2) the epsilon-copper precipitate has overaged and provides little strengthening. Tempering of the dislocation structures also will be experienced in the HAZ. These conditions would be expected to create a softened region in the HAZ; this is observed in some situations.

The occurrence of softened zones has been reported by numerous researchers (Jesseman and Schmid, November 1983; Kim and Choo, September 1988; Lundin et al., December 1990; Abe, Kurihara, and Tagawa, September 1988). Figure 7



(Source: Lundin et al., December 1990).

Figure 7. Typical microhardness distribution across heat-affected zones (HAZ) for HSLA-80/A710 at three different cooling times (800 °C to 500 °C).

shows the microhardness distributions across several HAZ in an HSLA-80 weldment (Lundin et al., December 1990). The longer cooling times imposed by the higher heat inputs tend to produce larger and softer soft zone regions. Jesseman and Schmid (November 1983) found the softened region became wider as the energy input was raised. However, a detrimental effect of such a softened region is not always present. Transverse tensile tests showed that this softened region did not affect the joint strength at heat inputs up to 125 kilojoules per inch (kJ/in.) on 2.25 in. thick plates. Recent work at Lehigh University has confirmed that low heat input weldments in thick section steels may not provide sufficient time for softening to occur. The softened zone is more apparent in thinner sections or at higher heat inputs. If the welded section is small enough it may be possible to soften the entire section by the heat of welding.

Heat Input Restrictions

The necessity for maximum heat input controls when welding some high strength quenched and tempered steels stems primarily from two motivations. It is desirable to avoid low toughness microstructures that develop in the slowly cooled HAZ. Also the low strength and toughness that can develop in weld metals deposited at high heat inputs should be avoided. Currently heat input restrictions are imposed by military standard 1689 for welding U.S. Navy ship structures. The requirements state that a maximum of 45 kJ/in. is permitted when welding HSLA-80 less than 1/2-in. thick, and a maximum of 55 kJ/in. is permitted when welding HSLA-80 1/2 in. thick or greater. The standard also allows the possibility of exemption from these requirements. Several shipyards have qualified procedures that allow extensions of these heat input limitations.

Some concern for limiting the heat input comes from the occurrence of local brittle zones (LBZs) that are known to occur in HSLA steels. Typically these LBZs are associated with the coarse-grained HAZ or intercritically reheated coarse-grained HAZs. Scoonover (November 1990) suggested that the typical heat inputs used in shipbuilding result in small, discontinuous LBZs. Probabilistic fracture mechanics models incorporating LBZs indicate only a small increase in the probability of service failure caused by the presence of LBZs (Scoonover, November 1990).

Suppliers

Four major domestic steel companies supply A710-type plate: Armco, Bethlehem, Lukens, and USX. All four supply principally the Grade A composition, most often in Class 3 condition. USX indicated they occasionally also have supplied some of the

low chromium grades (Grade B and C) to the oil drilling industry. Bethlehem supplies plate rolled from ingots furnished by Armco.

Recently USX announced the availability of seamless piping produced from a modified A710 composition. This modification adds approximately 0.015 percent titanium and eliminates the chromium. This composition apparently has increased the resistance to toughness degradation in the HAZ as well as lowered the sensitivity to reheat cracking (Kapadia et al., March 1993).

Only limited structural shapes are currently produced in any of the A710 compositions. The only commercially available shape known to the authors are rectangular tubular welded beams supplied by Bull Moose Tube Company. Ingalls Shipyard has evaluated built-up structural shapes for stiffeners for ship hulls. These shapes were "T" shapes welded from A710-type wrought strip.

No casting alloys have been successfully produced commercially in the HSLA type of composition. ESCO Foundry conducted a research project in the casting area under the auspices of the National Shipbuilding Research Program (Churchill, Devletian, and Singh, May 1991). ESCO Foundry discovered that the desired properties—excellent strength and toughness—possible in the wrought condition could not be developed in a casting of the low-carbon, copper age-hardenable type of compositions. ESCO Foundry has had some success in developing a nonage-hardening steel casting composition that approaches the combination of properties exhibited by the A710-type wrought alloys. There has been some application of an age-hardenable cast steel in cast nodes for offshore platforms (Stuart, January 1991). These are not copper age-hardening alloys but attain their properties through a double normalizing treatment and take advantage of vanadium and niobium additions to control grain size and provide some precipitation hardening. Castings have not replaced fabricated nodes on a large scale mainly because of the experience that already exists for fabricated nodes. The long delivery time that inhibits design changes also has limited the application of these castings.

Some work currently is underway at Lehigh University's Advanced Technology for Large Structural Systems (ATLSS) Center evaluating the feasibility and properties of some experimental A710-type castings for special beam-to-column connection assemblies. These castings have exhibited strengths similar to the wrought A710 materials with some degradation of the toughness properties. Nevertheless, the toughness is superior to the A572 beams and columns being connected.

3 Applications

U.S. Navy

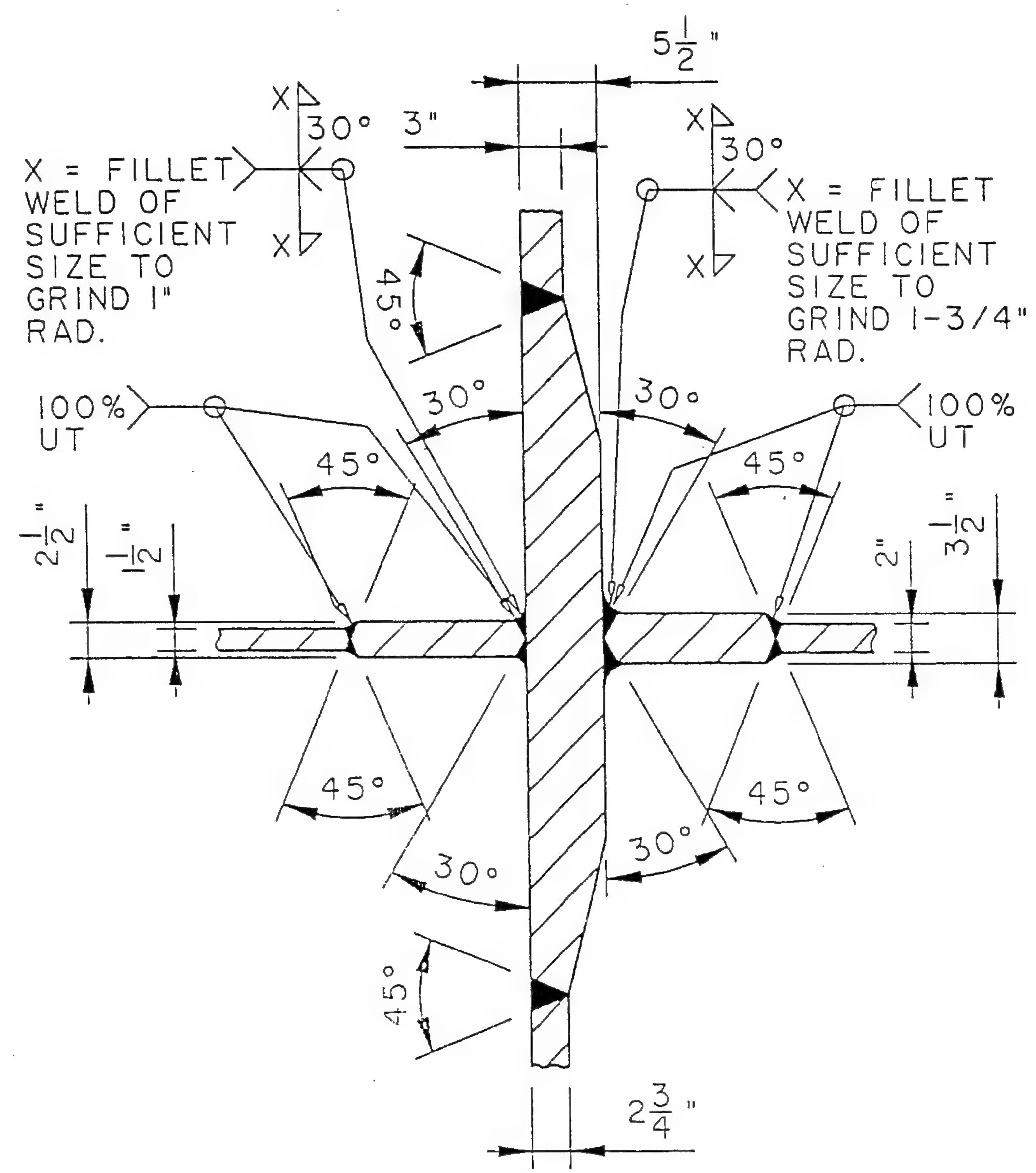
The U.S. Navy has conducted extensive evaluations of the advantages of HSLA steels (Kvidahl, July 1985); and broad application of these steels has occurred in the shipbuilding program. The prime benefit generated by HSLA steels is the ability to weld without preheat. Sustained preheat and interpass controls needed when welding high yield (HY)-80 and HY-100 steels cost nearly $1.5 million for a typical ship, and can cost up to $15 million for larger units. Thus significant savings accrue when the A710-type materials are substituted for the HY steels. Most of the HSLA tonnage used in the 80 grade material is as hull plate and is 3/4 in. thick or less. The 100 grade plate is used up to 3 in. thick, mainly in deck plating.

The Appendix is a bibliography of U.S. Navy reports (compiled by E. Czyryca of the Naval Surface Warfare Center) that deals with HSLA steels.

Offshore Drilling Equipment

Bethlehem-Beaumont fabricated a critical connection in a column leg-to-mat deck joint of an offshore oil platform jacking rig using 5.5-in. thick A736 plate (West, May 1987; Anderson, Hyatt, and West, September 1987). The plate was purchased in the quenched (not aged) condition. The lower yield strength of this condition permitted easier rolling to the required diameter. The subassembly was welded then precipitation hardened to attain the desired strength level. Figure 8 shows the connection detail for this rig. The steel was supplied by Kawasaki Steel at a delivered cost of $0.58/lb in 1986.

Ingalls Shipbuilding Division of Litton Industries has had significant experience with A710-type steels for construction of offshore oil drilling platforms (Kvidahl, May 1983). This experience was partly responsible for the U.S. Navy's introduction of A710-type steels in the construction of U.S. Navy cruisers (Kvidahl, July 1985). Recently USX introduced a new seamless pipe product for offshore tubular applications (Kapadia et al., March 1993).



The 5.5 in. thick section was fabricated from an A710-type alloy.
(Source: Anderson, Hyatt, and West, September 1987).

Figure 8. Joint detail of column leg-to-mat deck from offshore oil platform jacking rig.

H.R. Castner of the Edison Welding Institute recently reported on some weld development work using an A710-type material for tendons and riser pipe for what will be the world's deepest tension leg platform in 2,860 ft of water (Castner et al., April 1993). Three steel grades with nickel content varying from approximately 1 to 2 percent were evaluated over a range of heat inputs from 30 to 75 kJ/in., and preheat and interpass temperatures from 75 to 700 °F. Optimum procedures based on HAZ toughness results were found to be at heat inputs between 45 and 50 kJ/in., with preheats of 75 °F and maximum interpass temperatures of 375 °F. These procedures are based on welding 26-in. diameter pipe with a wall thickness of 1.3 in.

Mining Industry

Some heavy duty truck bodies designed for coal hauling have been fabricated with NI–COP®, the trade name for an A710-type alloy produced by Armco Steel Corp., Middleton, OH. Both the heavy box frame and bed of these off-road trucks were made from the A710-type steel. Some dredging equipment has used A710-type alloy fabrications. These steels were chosen because of requirements for high reliability and the ability to be successfully welded in cold, harsh environments.

Oil and Pipeline Applications

Grade C and B compositions are used in the oil industry in the form of electric resistance welded tube and pipe. Armco also reports that some 17 ton check valve assemblies used in the Alaskan pipeline have been manufactured from NI–COP®.

Water Systems

A recent fabrication of a large penstock specified an HSLA-80 material of the straight sections and an A736 (the pressure vessel grade of A710) 4-in. thick plate for some bifurcations in the line. Currently there is litigation concerning this fabrication, thus information about this application is not freely available. However, the thick sections were welded without preheat. Unfortunately, the high strength weld metal selected for this application apparently was not amenable to welding without preheat. Significant transverse weld metal cracking developed.

Application of HSLA steels to water and oil storage tanks has not yet occurred. Discussions with several fabricators indicate that there are possible needs for HSLA steels. Several large tanks recently were fabricated with 80 and 100 ksi yield strength plates (A537 Class 2 and A514). The use of the A710-type materials most likely would have provided lower fabrication costs in those instances. However, the owners, designers, and fabricators have been hesitant to use A710-type steels because they are not specifically permitted by the applicable design specifications of the American Water Works Association or the American Petroleum Institute.

4 Discussion

The A710-type materials provide an excellent combination of strength, toughness, and weldability unavailable in any conventional structural steel. Because these materials can be welded without preheat, there is a potential for significant fabrication cost savings. Because of the high strength, there is a potential for weight saving design applications. Because of the high toughness, new structures fabricated from A710-type materials will be more defect tolerant and will provide an inherently safer structure. Another benefit that can be expected is reduced repair cycles because of the improved weldability.

The foregoing benefits must be weighed against the higher costs of HSLA steels. Because of the additional heat treatments and higher alloy content, these steels are more expensive than conventional micro-alloyed structural steels. However, when used by the U.S. Navy, these HSLA materials provide significant savings and improved performance. But in this instance HSLA materials are being used to replace a quenched and tempered steel of higher alloy content and relatively poor weldability. The cost savings projected by the U.S. Navy will not occur if A710-type materials are used to replace conventional structural steels. The current price for A710-type steels ranges from $0.75 to $1.00 per pound; the price for conventional structural steels ranges from $0.45 to $0.65 per pound. If the A710-type steels are to replace conventional structural steels successfully, then there must be significant design, fabrication and operational benefits to justify the added material costs.

One factor that causes high strength materials to be less able to take full advantage of the higher strength is that the fatigue life in large, welded structures is principally a function of the weldment geometry rather than the base metal strength (Dexter, Fisher, and Beach, June 1993). This means any implementation of HSLA materials that attempts to use the benefit of the higher strength by reducing section size and weight, and at the same time raising the operating stress, will have reduced fatigue capacity. The fatigue S-N curves for A710-type large scale weld details are not significantly different than S-N curves from similar weld details fabricated from lower strength, carbon-manganese steels. To a certain degree, the fatigue capacity of these weld details may be improved by auxiliary treatments such as peening or gas tungsten arc remelting of weld toes (Fisher and Dexter, June 1993). Obviously

these extra steps increase the cost and lower the economic benefits of the use of HSLA material.

In many structures (e.g., ship hulls and storage tanks) if the allowable stress is increased to take advantage of high strength steel, compression stability is more likely to become a governing design consideration. If the design becomes overly thin, local buckling may occur at service load levels and lead to fatigue cracking from repeated out-of-plane deformations. Factors that influence buckling include residual stress, geometrical imperfections, and the steel elastic properties. Unfortunately, these factors are to a large degree independent of the steels' tensile strength. Therefore, design configurations and details with improved stiffness and fatigue life are required for efficient use of high strength steel.

One area that holds promise for the application of HSLA materials is in structural details that can significantly benefit from the superior toughness of these materials. Recently the feasibility of such a benefit was shown when an A710 was substituted for an A572 in small attachment tabs in a beam-to-column weak axis connection (Figure 9). When tested with the A572 connection, the joint failed prematurely due to fracture of one of the A572 connecting plates. The failure was attributed to unavoidable strain concentrations caused by this design detail. When tested with A710 connection tabs, this design showed a significant increase in connection capacity. The nondimensional load-deflection curves of the two tests are shown in Figure 10. The A572 test did not reach the plastic limit load (Vp) of the beam. The A710 test ultimate load exceeded Vp by about 20 percent and the connection showed good ductility. Final failure of the A710 connection was caused by a fracture of the beam tension flange; the failure of the A572 connection was in the top flange connection tab (Hettich, June 1990).

For structures designed for seismic applications, the steel must be able to develop fully plastic hinges and absorb energy. This behavior is limited by either exhaustion of ductility or, more likely, ductile tearing of the material, especially at the welded or bolted connections. Therefore, the superior ductility and toughness of the A710-type steels make them ideal for such uses. These qualities make the HSLA steel ideal for applications involving biaxial tension, e.g., gusset plates. Structural shapes, particularly "I" beams, rolled or fabricated from A710 steel could make excellent shear links in eccentrically braced frames (Figure 11a). These frames are gaining acceptance as a means of dissipating energy by cyclic plastic deformation of the short center shear link. Obviously, this use puts extreme demands on the material's ductility. Figure 11b shows knee bracing, which could be used to dissipate energy.

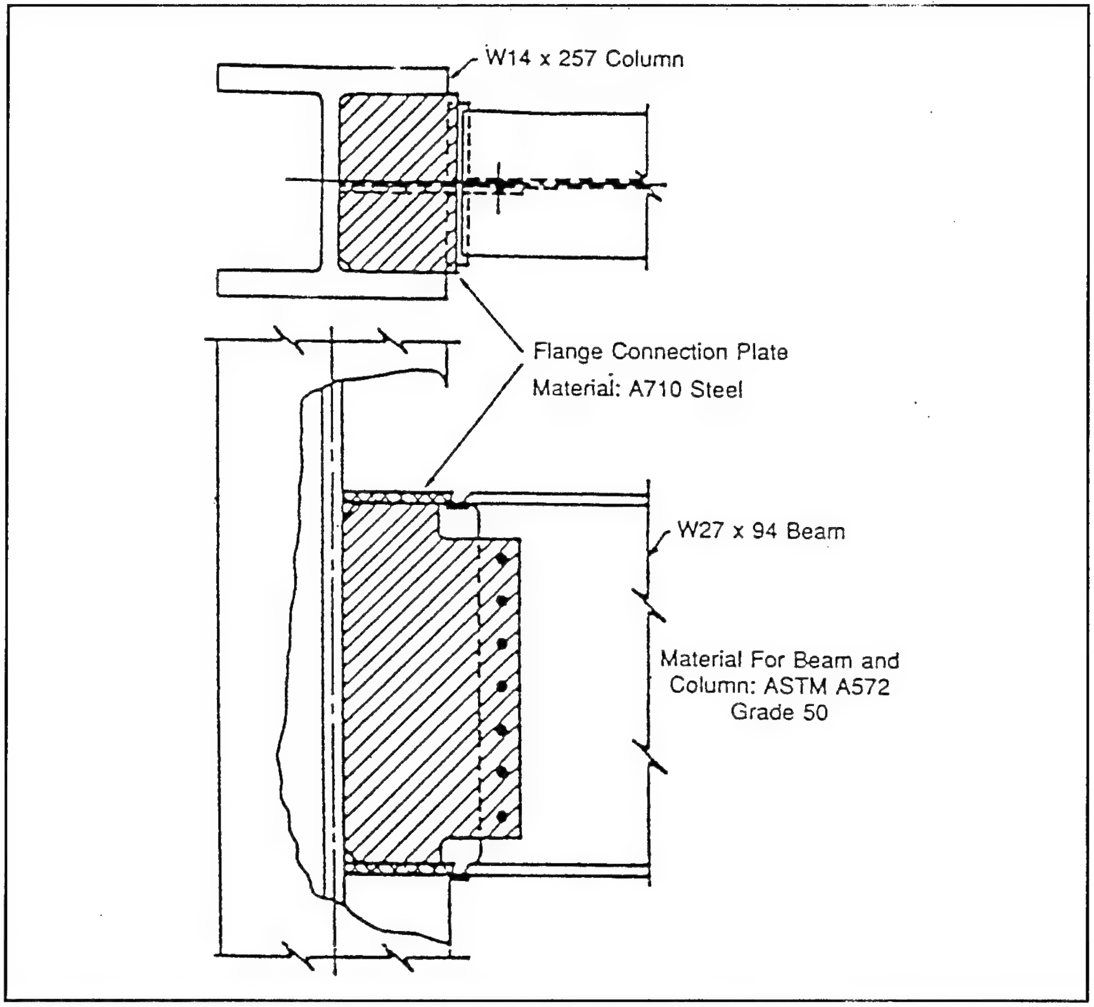


Figure 9. Beam-to-column weak axis connection test details with A710 steel.

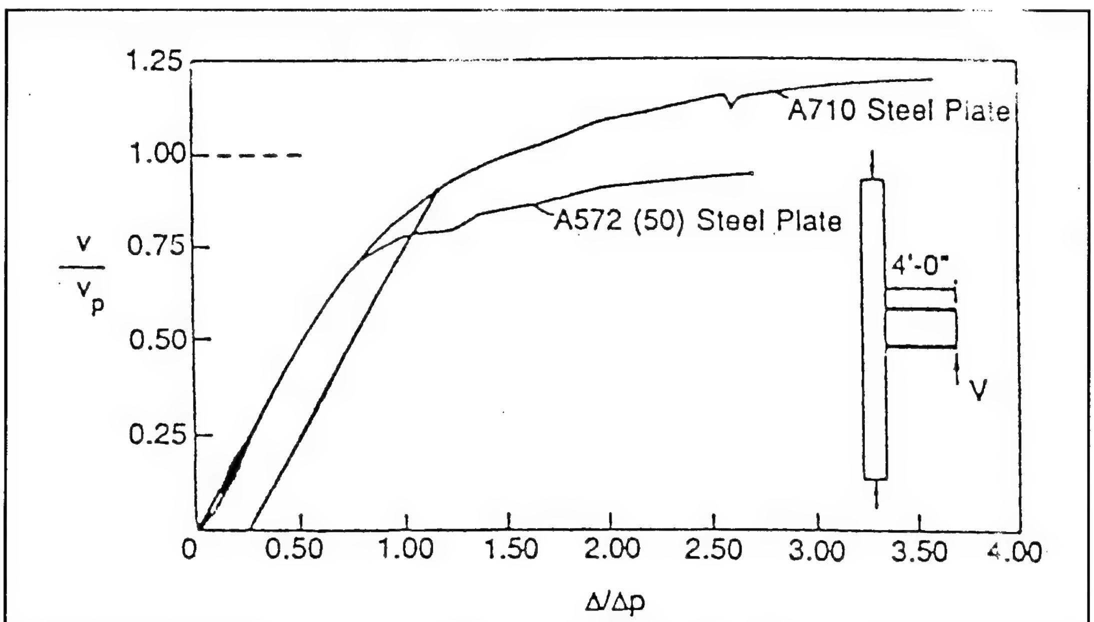


Figure 10. Load-deflection curves of beam-to-column weak axis connections.

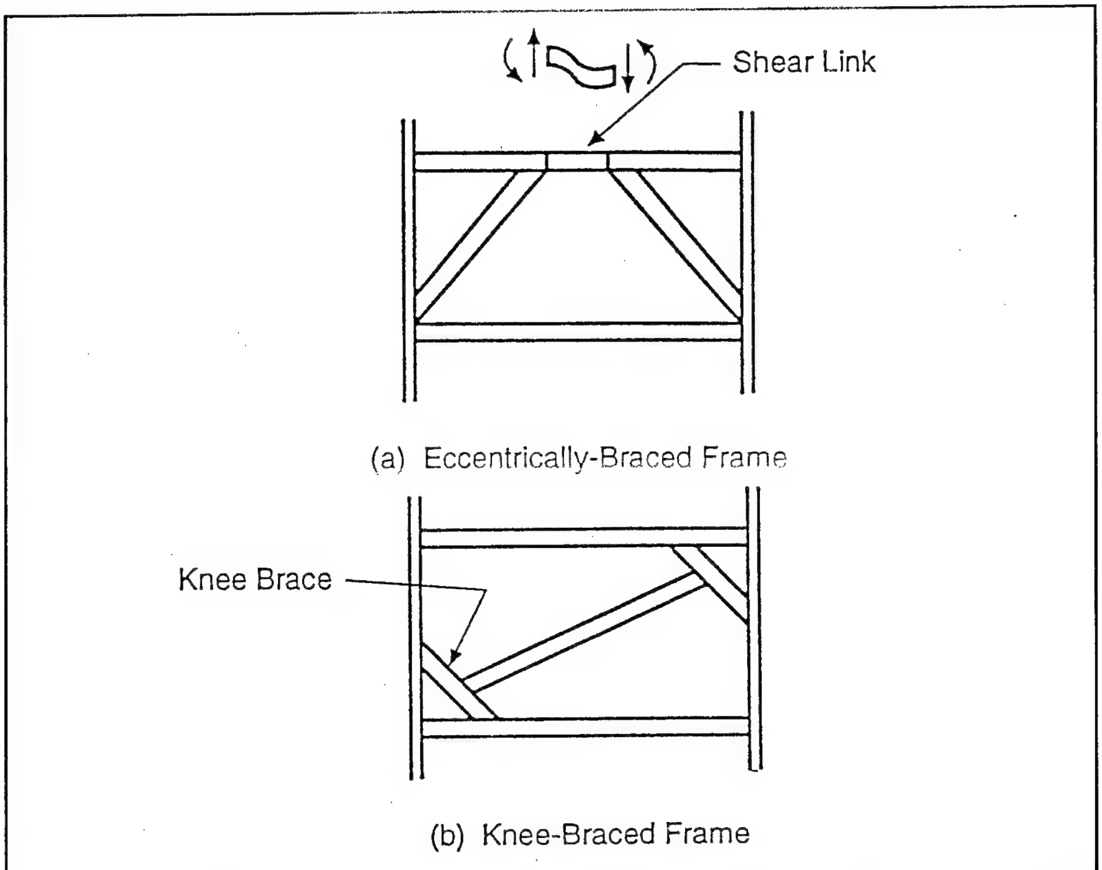


Figure 11. Applications for A710 high-performance steel which are required to undergo cyclic plastic deformation in an earthquake.

The improved weldability of HSLA steels should allow the use of thicker flanges in welded girders. This permits the use of members that are the same depth as conventional steel girders to achieve greater strength and stiffness. Thicker web and flange plates may reduce or eliminate the need for stiffeners in some instances. Elimination of the stiffeners, welds, and their distortion problems could improve the fatigue strength of the girder.

The excellent toughness and weldability of the HSLA steels should result in their applications involving repairs and retrofitting of bridges and other structures. Retrofitting may be done to improve resistance to fatigue or seismic loading. The use of these steels in repairs and retrofitting may be driven by the ability to weld in the field under poor conditions, possibly even from one side only. For example, the knee bracing shown in Figure 11b could be installed in existing frames using one-sided welding. High performance steel plates also are ideal for one-sided welding as doubler plates on existing structure. These doubler plates could be used to add capacity to or to repair damage caused by collision of vehicles or fatigue cracking. When repairing, if the original failure was due to loads higher than expected and the level of these loads cannot be changed, a replacement of the area or detail

in question could be made with a new A710 detail. Even if the original problem stemmed from fatigue, the increased flaw tolerance of the A710 steel will allow longer life.

Significant research activity designed to efficiently and effectively use the new high performance steels is developing. In 1991, the Civil Engineering Research Foundation (CERF, Washington, DC) recommended that top priority be placed on research and development to exploit the potential of high performance concrete and steel for the nation's infrastructure. In response, a planning committee made up of experts from industry, academia, and government are planning and formulating a high-performance construction materials program (COMAT). Industrial leadership for studies of high performance steel will be provided by the High-performance Steel Coordinating Committee to be organized by the American Iron and Steel Institute and the American Institute of Steel Construction. Required Federal support for COMAT is envisioned at $2 to $4 billion over a 10-year plan (CERF draft planning document, not dated).

A recently initiated Federal Highway Administration project (Project DTFH61-93-R-00007) focuses on the feasibility of using HSLA steels in highway bridges. Specifically, the research is supposed to investigate the potential for using the HSLA steels in present bridge design criteria, and then extending the research into new bridge design criteria.

New bridge designs should be the result of innovative concepts that make more efficient use of the high performance steels. The project team–composed of Modjeski and Masters, Inc., and the primary consultant, Lehigh University's ATLSS Center–is directing its efforts toward innovations through specification change and variations in the basic structural form. Another focus of this work will include attempts at developing designs that take advantage of the extraordinary fracture toughness of HSLA steels. And, because of the reduced need to preheat, field welding may be considerably more attractive to many designers than it has been in the past. The project team believes the ability to capitalize on the fracture toughness and weldability of the HSLA steels will be the key to finding "better ways to build safer, more efficient bridges....the bridge of tomorrow" (Modjeski and Masters, October 1992).

A significant difficulty in trying to take advantage of the ability to be welded without preheat is the lack of reliable weld metals with similar capabilities. Hydrogen-assisted cracking is sometimes experienced in the weld metal when the HSLA steels are welded without preheat.

Hydrogen-assisted cold cracking requires a tensile residual stress, a susceptible microstructure, and hydrogen. In a weldment the tensile residual stress results from the weld shrinkage strains and is proportional to the amount of restraint. The susceptible microstructure generally is accepted to be martensite. As mentioned previously, the principal microconstituent in the coarse grain HAZ in the A710 steel is bainite, thus resistance to hydrogen-assisted cracking is high. Hydrogen is commonly present in varying amounts in the atmosphere surrounding the arc; the amount that ends up in the HAZ depends on the quantity present as well as the transformation characteristics of the weld metal and base metal. When welding a typical structural steel (yield strength approximately 40 ksi), the weld metal typically is a low-carbon, relatively-lean alloy composition. As the molten pool moves along, the on-cooling transformation of the weld metal usually occurs before that of the base metal HAZ directly beneath. Thus, for a short distance, untransformed austenite is beneath the weld metal that has transformed, most typically to a ferrite-pearlite mixture (Figure 12). Because the solubility of hydrogen is higher in austenite than the transformed product and the diffusivity of hydrogen is faster in the transformed products, the hydrogen dissolved in the weld metal tends to get pumped across the fusion line into the HAZ.

The composition of the weld metal must be richer to match higher strengths. Wrought base metal can develop high strength though complex aging reactions and thermal-mechanical treatments, but the weld metal typically can develop strength only through alloy additions. PWHT can age the weld metal. However, PWHT generally is not used specifically for weld metal strengthening, and the procedure

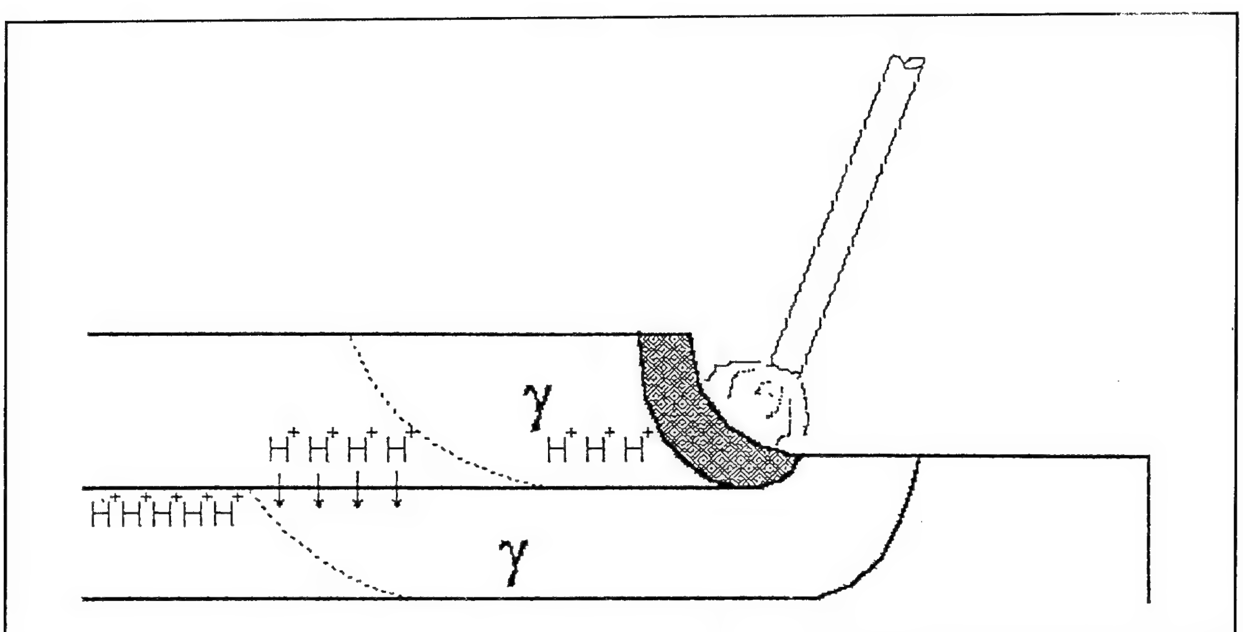


Figure 12. Schematic representation of hydrogen movement during weld deposition of typical structural steel.

frequently is impractical because of size restrictions, e.g., in large structures. In situations with extremely lean high strength alloys, the transformation characteristics of the weld metal-base metal combination may be as shown in Figure 13. Because of the higher alloy content of the weld metal, the dissolved hydrogen in the weld metal remains in the weld metal, and there is a tendency to pump hydrogen from the base metal into the weld metal.

When the A710-type materials are welded without preheat, in situations with high restraint and perhaps stress raisers, hydrogen-assisted cracking may be expected to occur in the weld metal. Weld metals in these higher strength ranges typically have not been used without preheat in the past. Additionally, the transformation characteristics of the typical base metals used with these weld metals have been such that the hydrogen was not driven into the weld metal as may be the case with the lean A710-type steels. In some instances the result is that cracking occurs in the weld metal. Improved weld metals are needed for full utilization of the no preheat benefit in A710-type materials and all the newer family of high-strength, low-carbon, low-alloy steels that develop strength via thermomechanical controlled processes.

There are two approaches to a solution to the high strength weld metal problem. The most straight forward is the development of new and better high strength weld metals that resist hydrogen cracking when welded without preheat. A second

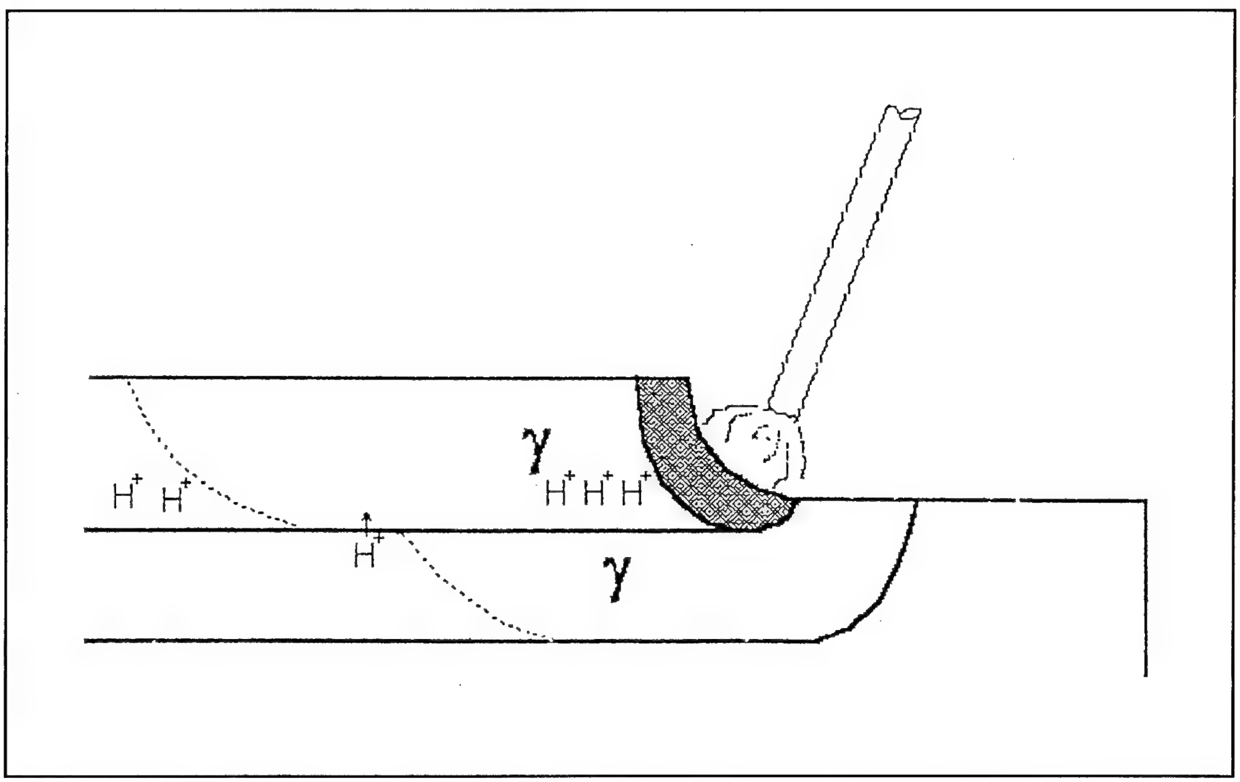


Figure 13. Schematic representation of hydrogen movement during weld deposition of new high performance low-carbon, lean-alloy steels.

approach–and perhaps a more near-term solution–is welding with undermatched weld metals. This is the current approach the U.S. Navy has taken on some of its HSLA-100 decking weldments. This approach generally requires a case-by-case evaluation in which the joint geometry and configuration are examined to determine the ability of the constraint offered to raise effectively the yield strength of the confined weld metal similar to the effect noted in braze joints.

Considerable work is ongoing to develop better high strength weld metal. Currently the National Center for Excellence in Metal Working Technology (NCEMT) in conjunction with the Naval Surface Warfare Center (formerly David Taylor Research Center) is directing a three-phase program to develop advanced welding consumables for high strength steels. The first phase, which involved participation from the Oregon Graduate Institute and the Ohio State University, evaluated 15 experimental compositions at two heat inputs. The results from these experiments have lead to the development of 10 additional wire compositions that are targeted to meet the program goals of strength and toughness. These experimental gas metal arc welding wires are being produced by Alloy Rods. Delivery of these wires has been made for Phase 2 testing of the weld metals at NCEMT. After Phase 2 testing, a decision will be made on the viability of a commercialization phase, Phase 3. Preliminary testing of the experimental weld metals on HSLA-100 at three different heat inputs and no preheating have shown no cracking.

Other weld metal development work sponsored by the National Shipbuilding Research Program is ongoing at the Colorado School of Mines. The objective of this project is the development of shielded metal arc electrodes for HSLA-100. Titanium and other microalloy additives are being used to produce desirable fine grain ferritic structure to modify the basic martensitic/bainitic metallurgy. This approach is aimed at meeting the yield strength and toughness requirements for HSLA-100 and to improve resistance to hydrogen-assisted cracking (Davis, November 1992).

The relatively severe sensitivity of the base materials to reheat cracking has lead to at least one major failure. Potential users of the HSLA steels are generally aware of this problem and in some instances appear to have become conservative in their approach to fabrication of these steels. The ability to refine the coarse grained HAZ, and thus significantly reduce the potential for reheat cracking, has lead some to consider the application of what is now termed controlled-deposition techniques for HSLA steel fabrication. There has been some concern that the sensitivity to reheat cracking could lead to interbead HAZ cracking as a result of the reheating of a coarse grained HAZ from subsequent passes; however, no failures have been attributed to this mechanism. The probability of this mechanism occurring is small unless associated with an interbead discontinuity such as lack of fusion. The application of a controlled deposition technique has been considered to avoid the possibility of interbead cracking. This approach appears to be overly conservative.

Controlled deposition requires a buttering of the base metal with a low heat input first layer followed by a higher heat input second layer with sufficient heat to refine the coarse grained HAZ of the first layer. Obviously this deposition technique requires significant additional attention and eliminates most of the economic incentive to utilize HSLA steels. The initiation of reheat cracking in the HAZ requires a stress raiser to initiate or trigger the cracking. The authors have never experienced reheat cracking in HSLA steels when normal butt joints with typical reinforcement are subject to PWHT. Only when stress raisers are present, does the coarse grained HAZ crack.

The occurrence of local brittle zones in the HAZ of the HSLA steels also has played some part in the hesitancy of fabricators to use A710-type materials. There is a discrepancy between the results of CVN and crack tip opening displacement (CTOD) tests of HAZs. Generally the CVN testing shows some toughness degradation, but not to unacceptable levels. The CTOD test, with a sharper notch and larger specimen, may sample more of the LBZ and thus reveal the larger toughness degradation of these zones. The question of how LBZ affects the fracture behavior of full-scale structures has not been answered adequately.

Corps construction, which includes water control structures and buildings, frequently is governed by needs for fatigue and earthquake-resistant designs. New construction would not benefit by the use of A710 steels because larger section thicknesses still are required for fatigue resistance. Retrofitting and repair projects may receive some benefit from the use of the A710 steels by strengthening the members and improving the fatigue and seismic strength. Corps projects will not benefit from the weldability characteritics of the A710 steels because the materials normally used have fairly good welding characteristics.

5 Conclusions and Recommendations

Corps of Engineers steel construction requires fatigue and seismic strength, but not necessarily high tensile strength. These requirements therefore rule out using ASTM A710 steels for typical new construction projects. Retrofit and repair applications could benefit from the use of A710 steel, but additional work on joint and connection designs would be required.

The HSLA steels have the potential to provide significant savings in fabrication costs as well as increased reliability and safety for high tensile strength applications. However, some applications should be avoided because of the metallurgical characteristics of the HSLA steels; specifically, PWHT should be avoided because of the reheat cracking susceptibility. Technology gaps need to be fully investigated so confident application of the HSLA steels can occur. Some of the work required to close these gaps is currently underway; additional work may be needed in some situations.

Metallurgically three problem areas have inhibited the application of the HSLA steels:

- the inability of currently available weld metals to match the weldability of the ASTM A710-type steels,
- the sensitivity of these materials to reheat cracking,
- the toughness degradation due to LBZ as measured by CTOD.

Several research initiatives currently are under way for the first problem area. Additional efforts in this area may not be required. No additional recommendations should be made until the direction the current programs are headed is clear. Most of the need for this work is required by the application of higher strength–100 ksi yield strength and above–type steels and is not as critical for the application of the 80 ksi A710 steel.

The need for additional work in the area of reheat cracking susceptibility is not a high priority at this time. The problem is not expected to be encountered in structural applications because no PWHT is required. The question of interbead HAZ cracking due to reheat cracking sensitivity does not appear significant based on the

fabrication experience to date. If reheat cracking occurs it would be associated with other welding discontinuities that would require removal and repair regardless of the reheat crack sensitivity.

The problem of LBZs and their effect on overall structural fracture has not been satisfactorily resolved, and there is some question about whether a problem exists. Newer A710 modifications appear to have improved the resistance to the apparent HAZ toughness degradation measured by CTOD (Scoonover, November 1990). There is a discrepancy between the results of CVN and CTOD testing. Full-scale testing of actual weldments will resolve this question. It is recommended that after a complete literature survey of this question, full-scale tests in conjunction with CVN and CTOD tests should be conducted to determine the effect of LBZ on the overall performance of a full-scale structure.

From the structural engineering standpoint, two areas require further examination to allow the application of the HSLA steels and utilize them to their maximum benefits. These are the buckling stability and the fatigue life of structures. Both properties are not improved by increases in yield strength and will be degraded if existing designs substitute thinner sections to take advantage of the higher strengths available with these materials. Novel approaches to designs for utilization of higher strength steels are needed. However, recommendations for additional work should be delayed until the extensive COMAT plan develops into its final format.

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Acronyms and Abbreviations

ATLSS	Advanced Technology for Large Structural Systems (Center for)
CCT	continuous cooling transformation
CERF	Civil Engineering Research Foundation
COMAT	construction materials program
CTOD	crack tip opening displacement
CVN	Charpy V-notch
ft·lb	foot pound
HAZ	heat-affected zone
HSLA	high-strength low-alloy
HY	high yield
kip	kilopound
kJ/in.	kilojoules per inch
ksi	kilopounds per square inch
LBZ	local brittle zones
MPa	megapascals
NCEMT	National Center for Excellence in Metal Working Technology
PWHT	post weld heat treatment
SRC	stress relief cracking
Vp	plastic limit load

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ATTN: CECW-P
ATTN: CECW-PR
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ATTN: CEMP-C
ATTN: CEMP-M
ATTN: CEMP-R
ATTN: CERD-C
ATTN: CERD-ZA
ATTN: CERD-L
ATTN: CERD-M
ATTN: DAEN-ZC
ATTN: DAIM-FDP

CECPW 22310-3862
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US Army Engr District
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US Army Engr Division
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INSCOM
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USA TACOM 48397-5000
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Defense Distribution Region East
ATTN: ASCE-WI 17070-5001

Defense Distribution Region West
ATTN: ASCW-WG 95296-0100

US Army Materiel Command (AMC)
Alexandria, VA 22333-0001
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FORSCOM
Forts Gillem & McPherson 30330
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TRADOC
Fort Monroe 23651
ATTN: ATBO-G

Fort Belvoir 22060
ATTN: CETEC-IM-T
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USA Natick RD&E Center 01760
ATTN: STRNC-DT
ATTN: AMSSC-S-IMI

US Army Materials Tech Lab
ATTN: SLCMT-DPW 02172

USARPAC 96858
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CEWES 39180
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USA AMCOM
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Fort Leonard Wood 65473
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US Army ARDEC 07806-5000
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Engr Societies Library
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Defense Nuclear Agency
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Defense Logistics Agency
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US Military Academy 10996
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ATTN: Facilities Engineer
ATTN: Geography & Envr Engrg

Naval Facilities Engr Command
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ATTN: Public Works Center (8)
ATTN: Naval Far East Support Center
ATTN: Naval Facilities Engr Service Center
93043-4328

Tyndall AFB 32403
ATTN: HQAFCESA/CES
ATTN: Engrg & Srvc Lab

USA TSARCOM 63120
ATTN: STSAS-F

American Public Works Assoc. 64104-1806

US Gov't Printing Office 20401
ATTN: Rec Sec/Deposit Sec (2)

Nat'l Institute of Standards & Tech
ATTN: Library 20899

Defense General Supply Center
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Defense Construction Supply Center
ATTN: DCSC-WI 43216-5000

Defense Tech Info Center 22060-6218
ATTN: DTIC-O (2)

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